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RELATION OF THE HIGH-TEMPERATURE TREATMENT OF HIGH-SPEED STEEL TO SECONDARY HARDENING AND RED HARDNESS

BY

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RELATION OF THE HIGH-TEMPERATURE TREATMENT OF HIGH-SPEED STEEL TO SECONDARY HARDENING AND RED HARDNESS

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I. INTRODUCTION

The metallography of high-speed tool steel presents certain anomalies, the explanation of which is not clear from the usual conceptions of the mechanism of hardening in simple carbon steels, as, for example, the familiar polyhedral structure of properly quenched high-speed steel, which is often called austenite, although its physical properties are largely those of martensite. The explanation of such anomalies and a better understanding of the fundamental nature of high-speed steel is becoming more and more important as the peak of its development is being reached. It is even probable that future improvements will be largely in the technique of treatment and the adjustment of compositions to meet special requirements, in which cases fundamentals are of highest importance.

From this angle the problem of developing high-speed steel is one of constitution and not composition. In spite of a wide variety of compositions, the general characteristics are very similar. The most effective method of attack is, therefore, through the study of a number of the physical properties of a high-speed steel and their correlation with the corresponding characteristics of simple carbon tool steel, the recognized reference standard. The relations for carbon steels between heat treatment, microstructure, and physical properties have been rather thoroughly studied and are summarized in another paper.[1]

[1] Scott and Movius, see forthcoming paper, Thermal and Physical Changes Accompanying the Heating of Hardened Carbon Steels.

To make the proposed correlation, the microstructure, hardness, density, magnetic properties, and thermal characteristics of a typical high-speed steel as affected by various treatments were studied. The variety of properties investigated necessitated the cooperation of several other laboratories of the Bureau of Standards; H. S. Rawdon prepared the micrographs, E. L. Peffer made the density determinations, R. L. Sanford the magnetic tests, F. H. Tucker the chemical analyses, Miss H. G. Movius prepared the thermal curves, and several assistants aided in the work.

The previously published researches related to the present one are those of Edwards and Kikkawa,[2] Carpenter,[3] Yatsevich,[4] Mathews,[5] and Andrew and Green.[6]

Edwards and Kikkawa determined the effect of tempering temperature on the Brinell hardness of two series of chrome-tungsten steels of constant carbon content, chromium being variable in one series and tungsten in the other. A constant hardening temperature of 1350° C was used with one exception, and the density changes of one representative composition were determined. This work represents the most important contribution to date to the study of the physical properties of high-speed steels.

Carpenter investigated the effect of quenching and tempering temperatures on the structure and etching time of some high-speed steels.

Carpenter, Yatsevich, and Andrew and Green studied the critical ranges of high-speed steels as affected by maximum temperature and rate of cooling.

Mathews summarized his wide experience with cutting and physical tests of high-speed steel and reviewed developments since the classic experiments of Taylor and White.

These papers represent work on high-speed steel of a wide variety of compositions and treatments, so that it is very difficult to establish any relation between the various properties studied. The present work is, therefore, confined to one representative type of modern high-speed steel.

II. PHYSICAL PROPERTIES AND STRUCTURE OF HIGH-SPEED STEEL

For the purpose of this investigation tests were made on a standard type of high-speed steel, chosen because it shows sec-

[2] Edwards and Kikkawa, Jour. Iron and Steel Inst., **92**, p. 6; 1915.
[3] Carpenter, Jour. Iron and Steel Inst., **67**, p. 433, 1905; **71**, p. 377, 1906.
[4] Yatsevich, Rev. de Met., **15**, p. 65; 1918.
[5] Mathews, Proc. A. S. T. M., **19**, p. 141; 1919.
[6] Andrew and Green, Jour. Iron and Steel Inst., **99**, p. 305; 1919.

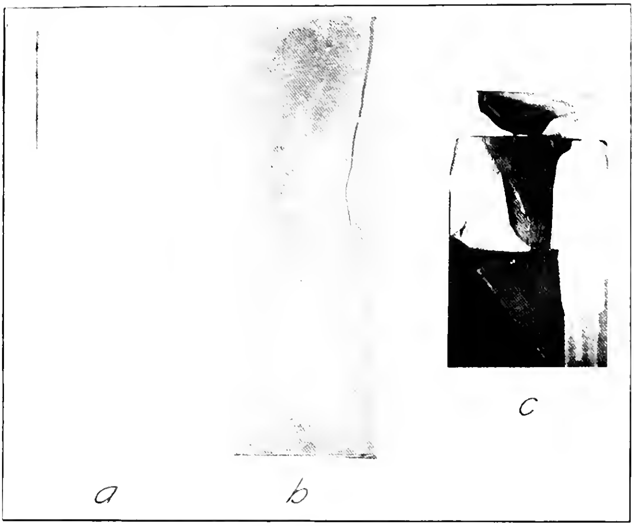

FIG. 1.—*Hardening cracks in high-speed steel*

Specimens of steel B, 1 by 5/8 by 3 inches, were hardened as follows; *a*, Quenched in oil from 1030° C, surface of fracture parallel to 1 by 3 inches surface, *b*, cooled in air from 1050° C, cracks perpendicular to 1 by 3 inches surface, *c*, quenched in oil from 1050° C, fracture parallel and at 45° to 1 by 3 inches surface

ondary hardening definitely. Unfortunately, sufficient material of one composition was not available, so that another composition was used also, this being purchased to duplicate the first as nearly as possible. As seen from the chemical composition given in Table 1, the two steels are very similar except in carbon content, which is low in the B steel for the brand used. As pointed out later, the lower carbon content, apparently, is responsible for a poorer steel.

TABLE 1.—Results of Chemical Analyses of High-Speed Steels

No.	C	Mn	Si	W	Cr	V	P	S
	Per cent	Per cent	Per cent	Per cent	Per cent	Per cent	Per cent	Per cent
A	0.77	0.25	0.47	17.8	3.5	0.74	0.020	0.03
B	.65	.31	.17	17.6	3.4	.73	.004	.04

For hardening, the specimens were placed in an electrically heated alundum tube furnace in which charcoal or illuminating gas was burned to prevent excessive oxidation. The specimens were slowly brought up to temperature and held there 15 minutes before quenching. The quenching medium was a light mineral oil. Tempering consisted of heating for 15 minutes in an oil bath for temperatures up to 250° C, in a nitrate bath up to 600° C, and in a chloride bath for higher temperatures. The use of these baths assured a uniform temperature quickly reached. All high-temperature measurements were made with platinum thermocouples.

The specimens were cut from a 1-inch square bar (steel A), and a 1 by 5/8 inch bar (steel B), both factory annealed. The magnetic test specimens were of 1 cm square cross section, the hardness specimens were of 1 by 1/4 inch (steel A), and 1 by 5/8 inch (steel B) section, and the density specimens about 1 by 1 by 1/2 inch (Fig. 7) and about one-half inch face cubes (Fig. 5). Specimens for micrographs were taken from the ends of the hardened magnetic test pieces.

In hardening, both steels usually cracked when quenched from the region of 1050° C. The cracks followed the contour of the specimens, thus indicating that they were characteristic of the steel and not due to inclusions. The cracking was particularly severe in the case of the low-carbon steel B, photographs of typical samples of which are shown in Fig. 1 (*a*, *b*, and *c*). This steel cracked also on cooling in air from 1050° C. It was, how-

ever, possible to obtain specimens on which the physical measurements could be satisfactorily made in spite of the cracks.

The density measurements were made by the usual method of weighing in air and in water, the specimens being dipped in alcohol prior to immersion in water to insure the absence of bubbles. The magnetic measurements were made in a long solenoid, corrections being made by the use of shearing curves. Other tests were made by the usual standard methods.

1. EFFECT OF QUENCHING TEMPERATURE

It has long been recognized that raising the quenching temperature increases the cutting efficiency of a high-speed steel

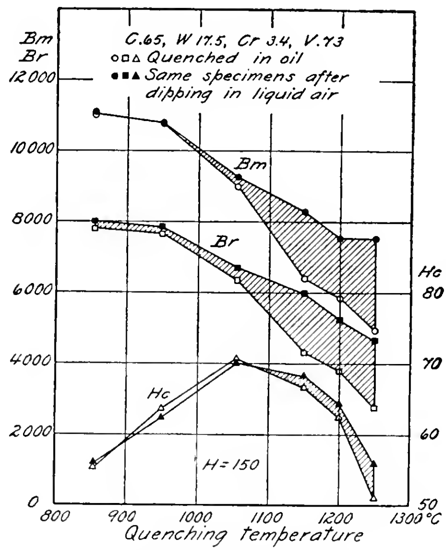

FIG. 4.—*Relation of quenching temperature and subsequent treatment in liquid air to maximum induction, residual induction, and coercive force of steel B*

tool, so that the highest temperature short of fusion is the best. Observations were therefore made to determine the effect of quenching temperature on the properties under consideration to obtain evidence as to the nature of the constitutional changes.

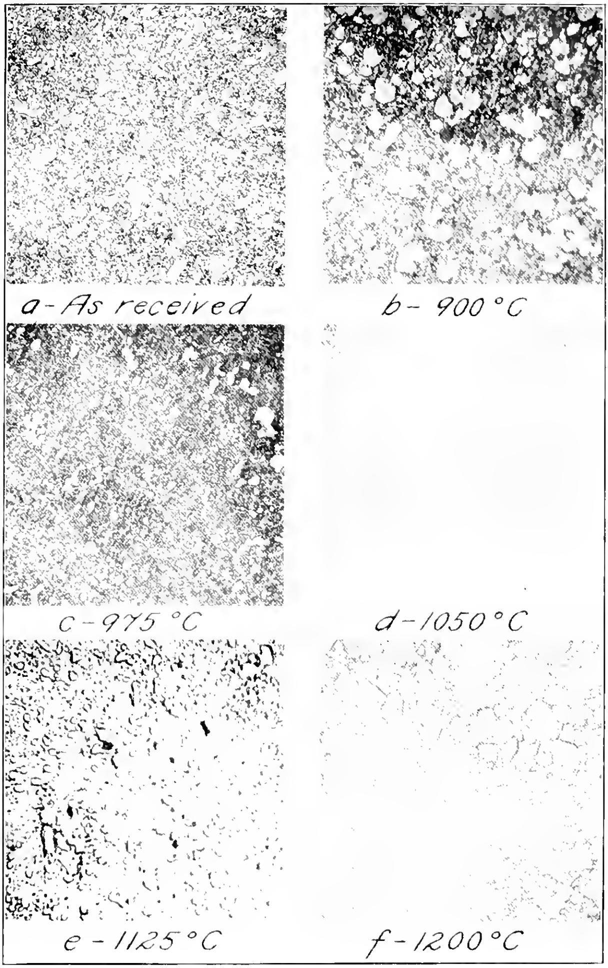

FIG. 2.—*Microstructure of specimens of steel A quenched from temperatures noted.* ×500. *Etched with 2 per cent alcoholic HNO_3*

a, Annealed, (*b*), 900° C, (*c*), 975° C, (*d*), 1050° C, *e*, 1125° C, *f*, 1200° C

FIG. 3.—*Microstructure of specimens of steel A quenched from temperatures noted. × 500. Etched with 2 per cent alcoholic HNO_3*

g, 1220° C; *h*, 1290° C, interior; *i*, 1290° C, structure near surface

The microstructure of steel A as quenched from several temperatures is shown in the micrographs of Figs. 2 and 3, the magnetic properties in Fig. 4 (lower curves of shaded areas), and the density in Fig. 5. The effect of quenching temperature on the Brinell and scleroscope hardness (recording instrument) of steel B is also shown in Fig. 5.

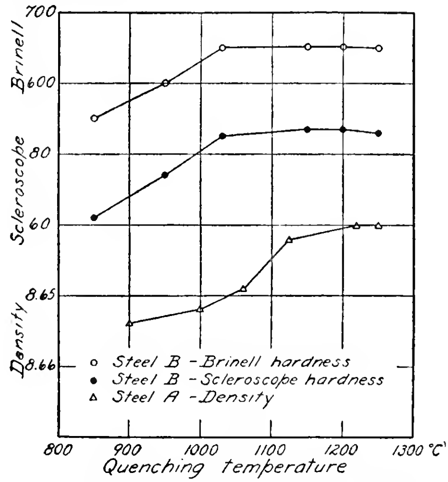

FIG. 5.—*Relation of quenching temperature to Brinell and scleroscope hardness of steel B and to density of steel A*

These data permit of a classification of the quenched specimens into two groups according to the nature of their physical characteristics. The properties of the first group, quenched from temperatures up to about 1100° C, vary in a manner distinct from those of the second group, quenched from above that temperature. Besides the change in slope of the curves, the physical properites of the two groups are affected differently on cooling below ordinary temperatures. Thus the magnetic properties of the specimens quenched from the high temperatures are markedly increased (Fig. 4, upper curve of shaded area) by immersion in liquid air, while the specimens quenched from the low temperatures remain practically unchanged. This is indicative of a constitutional difference, other than a continuously changing one, between the specimens of the two groups and has an important bearing on the anomalies of high-speed steel.

To distinguish between these two groups, specimens quenched from the lower temperature range will be referred to as given the low-temperature treatment, and those from the upper range as given the high-temperature treatment.

The microstructure of the samples given a low-temperature quench is obscured by the excessive amount of free carbide imbedded in the apparently structureless matrix. The microstructure of the specimens of this series will be called martensite from their physical characteristics, but it must be recognized that this may be a misuse of the term, depending, of course, on its definition. The specimens quenched from the high-temperature range show well-defined grain boundaries in a structureless matrix containing little free carbide. This structure is typical of properly quenched high-speed steel and will be called polyhedral.

The polyhedral structure, smaller volume change, constancy of hardness, and more rapid loss in magnetization with quenching temperature in this range are all suggestive of austenitization, though the polyhedral structure is not necessarily proof of it. It is seen by extrapolation of the magnetic properties that zero magnetization, and hence complete austenitization, would be attained for a quenching of about 1450° C if this temperature could be reached without fusion. That partial austenitization has occurred on quenching from the high-temperature range is shown, however, by the changes in physical properties on cooling below ordinary temperatures. This treatment, if carried to a low enough temperature, completes the A_3 transformation with a corresponding change in physical properties, direct evidence of previous austenitization. The effect on the magnetic properties of immersion in liquid air is shown in Fig. 4. The volume also increases on cooling below ordinary temperatures, a drop in density of 0.059 g/cm^3 being observed when a specimen of steel A, quenched from 1300° C, was cooled to −45° C. It must, therefore, be concluded that the specimens quenched from the high-temperature range are constitutionally different from those given the low-temperature treatment in that in the former case the steel is partially austenitic, but is not in the latter.

This characteristic of high-speed steel may appear peculiar to it, but on reference to the work of Maurer[7] one will find that more or less partial austenitization is common to simple high-carbon steel quenched from a high temperature, the degree depending, of course, on the carbon content and the temperature. This

[7] Maurer, Rev. de Met., 5, p. 711; 1908.

phenomenon is revealed by the change in density and in other physical properties on immersion of the steel in liquid air.

A further analogy between carbon and high-speed steel may be seen by comparison of the effect of the quenching temperature on the magnetic properties of high-speed steel (Fig. 4), with the effect of the same variable on those of a carbon steel (Fig. 6).

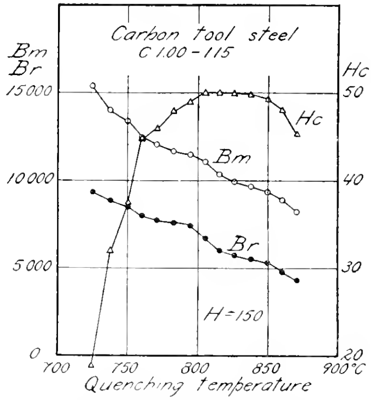

Fig. 6.—*Relation of quenching temperature to coercive force, maximum induction, and residual induction of carbon tool steel.* (Gebert)

The data for the carbon steel were taken from Gebert.[8] The similarity of these two figures is striking when the lack of any similarity in microstructure is considered.

2. EFFECT OF TEMPERING TEMPERATURE

The most interesting feature of the tempering of high-speed steel is the so-called "secondary hardening," which is revealed as an increase in hardness over the original (or a previous minimum) of certain high-speed steels given the high-temperature treatment and tempered in the neighborhood of 600° C. This, at first sight and in view of its absence in the usual carbon tool steels, is often considered a mysterious phenomenon. However, when the original condition of partial austenitization resulting from the high-temperature treatment is considered, the phenomenon appears quite natural.

[8] Gebert, Proc. A. S. T. M., 19, Part II; p. 117, 1919.

The effect of increasing the quenching temperature is to increase the amount of dissolved carbide which lowers Ar″ (Ar_{3-2} of martensitic steels, which is long and continuous for high-alloy contents) progressively, until, for a quenching temperature of about 1100° C, its end reaches room temperature. Any further increment of the quenching temperature—that is, quenching from the high-temperature treatment range—will cause the end of Ar″ to fall below ordinary temperatures, with the result of partial austenitization already noted. This phenomenon is analogous to the lowering of Ar_3 in iron-nickel alloys by increasing the nickel content, the essential difference being that in the preceding case the composition of the matrix can be changed by tempering, but it can not be so changed in the latter.

From the foregoing analysis it is evident that on tempering the partially austenitic, and consequently somewhat softened, steel, the dissolved carbide of the matrix will be gradually precipitated until a stage is reached at which Ar″ is no longer stable for the then

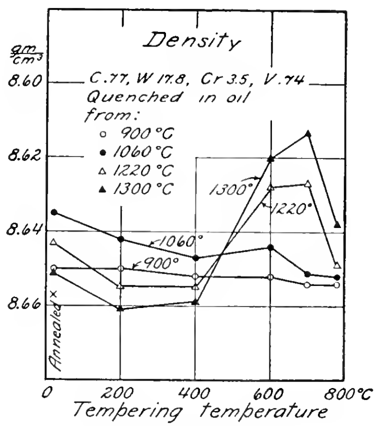

FIG. 7.—*Change in density of steel A with tempering temperature*

existing temperature and composition of the matrix. Ar″ will then complete itself as in the case of the austenitic carbon steel discussed in another paper.[9] The consummation of Ar″ implies martensitization, and consequently an increase in hardness, which the physical property and microstructure data verify.

[9] See footnote 1.

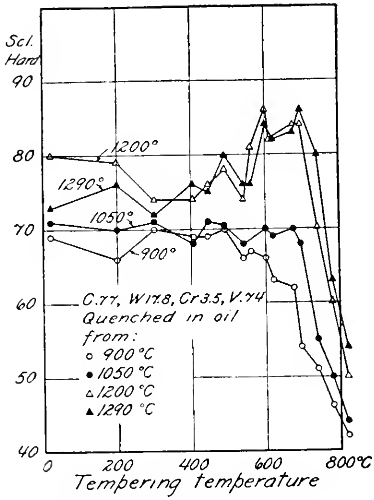

FIG. 8.—*Change in scleroscope hardness of steel A with tempering temperature*

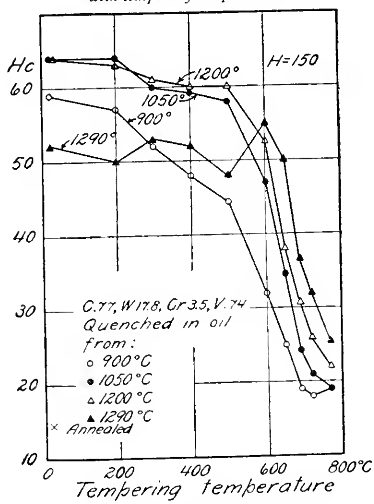

FIG. 9.—*Change in coercive force of steel A with temperimg temperature*

The effect of tempering temperature on the physical properties of steel A quenched from several temperatures is shown in Figs. 7 (density), 8 (scleroscope hardness), 9 (coercive force), 10 (maximum induction), and 11 (residual induction). The reciprocals of the density values have been plotted to show more clearly their

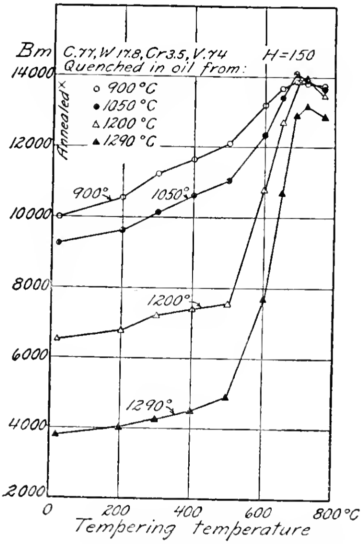

FIG. 10.—*Change in maximum induction of steel A with tempering temperature*

relation to the hardness values. In Fig. 12 are given Brinell and scleroscope curves for steel B.

The hardness curves (Figs. 8 and 12) show the phenomenon of secondary hardening only when the steel was quenched from the high-temperature range, above 1100° C, thereby confirming its relation to partial austenitization noted only in specimens quenched in the same temperature range. The curves representing the results of the low-temperature treatment are similar to those of carbon steels with the exception that tempering occurs at a much higher temperature, thus shortening the temperature range over which troostite is stable, with the result that the change from martensite to sorbite, or complete softening, is quite

abrupt. The density change is very small, indicating a slight solubility of the carbide in the low-temperature treatment range.

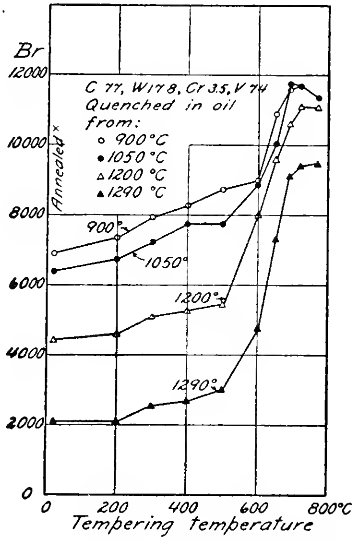

FIG. 11.—*Change in residual induction of steel A with tempering temperature*

Of chief interest is the tempering of specimens given the high-temperature—that is, the most efficient—hardening treatment. The density curves (Fig. 7) show very markedly a peak between 600 and 700° C, which is parallel to the hardness curves and corresponds to a large increase in volume. This is direct evidence of martensitization, and the magnetic properties offer further confirmation. The increase in maximum and residual induction (Figs. 10 and 11), normally indicative of softening, accompanies the rise in hardness in the secondary range.

Secondary hardening, as defined here, has been looked upon as a feature peculiar to high-speed steel. The work of Maurer, however, shows that hypereutectoid carbon steels quenched from the neighborhood of A_{cem} exhibit the same phenomenon, as revealed by density measurements, to a degree dependent on the amount of carbide retained in solution. The only difference between the high-carbon and the high-speed steel is that the peak comes at a

much lower temperature in the former case, and the tempering below the peak is naturally more evident. The increase in intensity of the peak with an increasing degree of austenitization is evidence of a general relation between secondary hardening and partial austenitization.

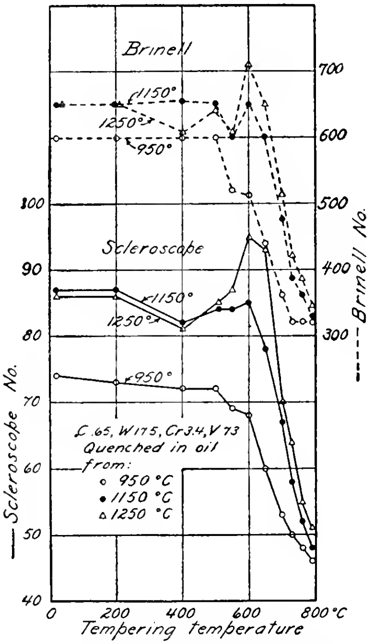

FIG. 12.—*Change in Brinell and scleroscope hardness of steel B with tempering temperature*

The conclusions arrived at from an examination of the physical changes on tempering as related to the phenomenon of secondary hardening should be capable of verification by observation of the accompanying microstructural changes. Micrographs of specimens of steel A quenched from 900, 1050, 1200, and 1290° C are given in Figs. 13, 14, 15, and 16, respectively, as tempered at several temperatures. The first two figures show the structure resulting from a low-temperature treatment and the second two from a high one.

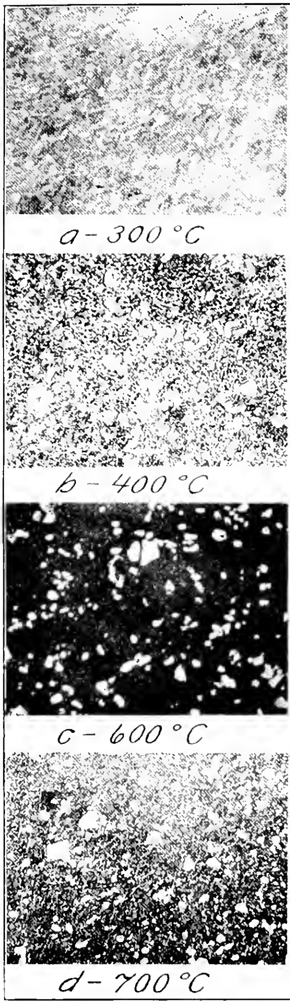

FIG. 13.—*Microstructure of steel A quenched in oil from 900° C and tempered as noted.* ×500. *Etched with 2 per cent alcoholic HNO_3*

a, 300° C; *b*, 400° C; *c*, 600° C; *d*, 700° C

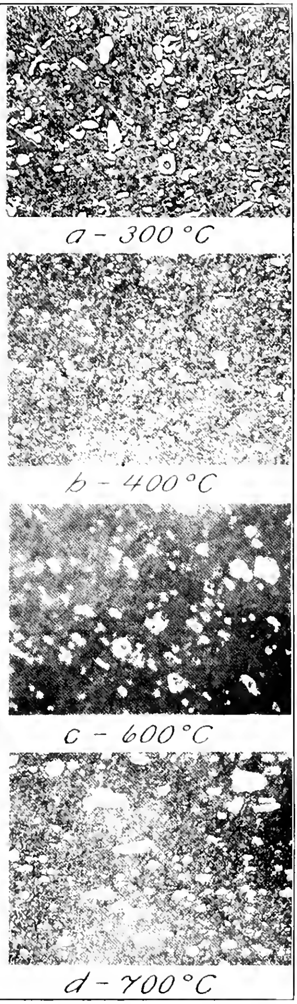

FIG. 14.—*Microstructure of steel A quenched in oil from 1050° C and tempered as noted.* ×500. *Etched with 2 per cent alcoholic HNO_3*

a, 300° C; *b*, 400° C; *c*, 600° C; *d*, 700° C

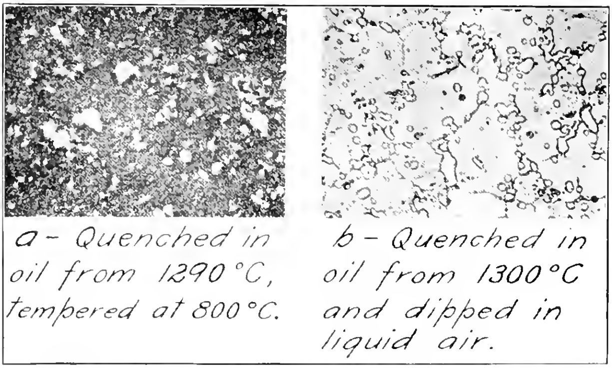

FIG. 17.—*Microstructure of steel A. × 500. Etched with 2 per cent alcoholic* HNO_3

a, Quenched in oil from 1290° C and tempered at 800° C; *b*, quenched in oil from 1300° C and dipped in liquid air

Micrographs *a* and *b* of Figs. 13 and 14 (low-temperature treatment) show an immature martensitic pattern resulting from tempering at 300 and 400° C. At 600° C the structure is apparently that of homogeneous troostite, and at 700° C the decomposition of troostite into sorbite is well advanced. It may be noted on referring to the relations between physical properties and microstructure of carbon steels [10] that the same general relations exist for the high-speed steel quenched from the low-temperature range.

The micrographs of specimens given the high-temperature treatment (Figs. 15 and 16) show a rather astonishing phenomenon. Tempered at 200 and 400° C (micrographs *a* and *b*), a well-developed, needle-like pattern, suggestive of martensite, is produced. When it is considered that the steel is still partially austenitic, the propriety of calling this constituent martensite is questionable. When tempered at 600° C (micrograph *c*), where secondary hardening first appears, a definite martensitic pattern quite similar to that of martensitic carbon tool steel is developed. For a tempering temperature of 700° C (micrograph *d*), the structure is that of the first stages of troostite (of tempering), and for a temperature of 800° C (Fig. 17, *a*) it is that of sorbite, the structure here being practically identical for all quenching temperatures and similar to that of the annealed steel.

The microscopic evidence is positive and confirmatory of the physical, namely, that the constituent accompanying the appearance of secondary hardening is martensite. The nomenclature of the patterns developed at 200 and 400° C must, however, await a more precise definition of the constituent martensite.

From the relations pointed out in a previous paper [10] between the heat evolution (Ac_t) observed on heating hardened carbon steels and the accompanying changes in physical properties and microstructure, it might be supposed that similar relations exist for the high-speed steel. Heating curves were, therefore, taken (Fig. 18) of steel A, quenched from three temperatures. The inverse-rate method was used, and the thermal characteristics noted on the curves are given in Table 2. There is evidence of a slight heat evolution (Ac_t) between 600 and 680° C, but it is not sufficiently intense to allow of any positive conclusions. This is probably due to the rather limited solubility of the carbide in the matrix and the sluggishness of its precipitation, the rapid heating rate required by thermal analysis allowing insufficient time for its consummation (note the loss in intensity of Ac_1 with increased quenching temperature).

[10] See footnote 1.

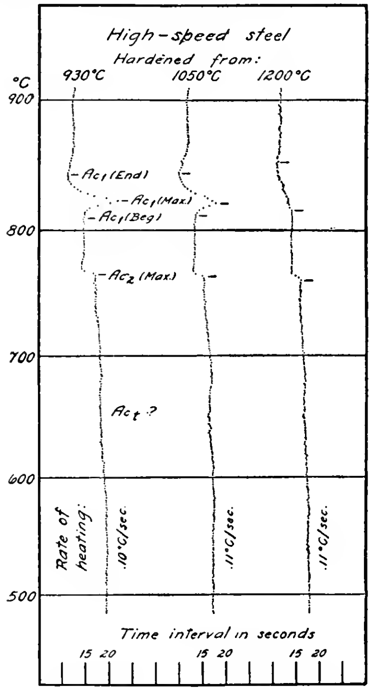

FIG. 18.—*Heating curves of quenched specimens, steel A*

TABLE 2.—Transformation Characteristics of Steel A on Heating Following Quenching

Quenching temperature	Rate of heating	Ac_2 maximum	Ac_1		
			Beginning	Maximum	End
°C	° C/sec.	°C	°C	°C	°C
930	0.10	765	810	823	843
1050	.11	763	811	821	844
1200	.11	760	....	815	852

These curves are in substantial agreement with Carpenter's [11] observations from differential curves on a tungsten-chromium high-speed steel, with the exception that he does not recognize the transformation Ac_2, this due to the lack of its characteristic features in differential curves.

[11] See footnote 3.

III. SIGNIFICANCE OF THE PHYSICAL CHARACTERISTICS OF HIGH-SPEED STEEL

Inasmuch as the high-grade high-speed steels have much in common in spite of the wide range of compositions represented, it is permissible to generalize somewhat from the observations made on the familiar type of high-speed steel used here. It must be remembered, however, that the degree of secondary hardening is a function of the carbon and chromium content and also, probably to a less degree, of the other alloying elements.

The rationale of the high-temperature treatment becomes evident from an analysis of the physical characteristics. By defining red hardness as the resistance to softening by tempering, one may see from the hardness versus tempering-temperature curves that the red hardness increases slightly with quenching temperature. This is evidently one potent reason for a high quenching temperature. The initial hardness also increases with the quenching temperature, at least in the low-temperature range, and this is a further highly desirable characteristic. With the increase in hardness there is a corresponding increase in volume and consequently in volume change on quenching, conditions which favor the formation of cracks. If, however, the steel be hardened from the high-temperature range with partial austenitization, the density change is not increased, and the steel, even if not less hard, is less brittle, thus counteracting its tendency to crack and furnishing the most satisfactory combination of properties.

There has been much discussion as to the value of tempering for maximum hardness or secondary hardening. This is evidently largely a matter of composition for a given treatment, as this factor determines the degree of secondary hardening possible. For a steel showing this phenomenon definitely, the increased hardness, if not accompanied by greater brittleness, which hardly occurs in this case, is certainly of value in a tool, no matter what its use. The increase in volume on tempering for secondary hardening deserves some consideration on selecting a treatment for a given tool. It is apparent that heating in service may change the dimensions of an untempered tool to a troublesome degree. Tempering for secondary hardening is, therefore, of general advantage, but it is certainly not of any value to temper at a lower temperature where these advantages do not accrue and where practically the same degree of effort is expended.

The comparatively long temperature range, about 100° C, in which secondary hardening may be obtained suggests that a considerable constitutional difference between the product of the

high and the low end of the range may exist, probably corresponding respectively to martensite and troostite. This being the case, it should be of interest to determine whether the low or high temperatures giving the same degree of hardness will give the better cutting results.

The sharpness of the changes in the magnetic properties and in the density would indicate that these properties furnish a valuable test for determining without destruction whether a tool has been properly tempered or not. Of the hardness tests, the Brinell is of little value on account of the very high degree of hardness involved. The scleroscope, however, if properly handled, is quite useful, although it is not as sensitive as the Brinell to the advancement of the change from martensite to troostite. On the whole the physical properties determined for a hardened steel, while of course not furnishing a direct criterion of its cutting efficiency, do offer a valuable indication of the constitution of a given steel, and this is the principal value of most physical and mechanical tests.

IV. SUMMARY

Attention is called to the importance of fundamental research applied to high-speed steel and the value of physical tests for this purpose.

The effect of heat treatment on the density, hardness, microstructure, magnetic properties, and thermal characteristics of a standard brand of high-speed steel was determined. The interpretation of these data permits the following conclusions:

1. A high-speed steel susceptible to secondary hardening is partially austenitic when quenched from a temperature high enough to produce this phenomenon.

2. The microstructure of steels hardened and tempered above 200° C is similar to that of carbon steels, although the same nomenclature in certain cases is not permissible.

3. The behavior of the physical properties of high-speed steel on heat treatment is analogous to that of hypereutectoid carbon steel.

4. The following reasons are given for the use of the high heat treatment: (*a*) Increase of red hardness; (*b*) increase of initial hardness; and (*c*) reduction of brittleness.

5. High-speed steel should preferably be tempered for secondary hardness.

WASHINGTON, April 27, 1920.

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